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TECHNICAL REPORT

WVT-11-6215

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A STUDY OF WELD HEAT-AFFECTED ZONES IN THE TITANIUM - 6Al-6V-2Sn ALLOY

BY

RICHARD E. LEWIS

KEH-CHANG WU

SEPTEMBER 1962

WVT-11-6215

U.S. ARMY WEAPONS COMMAND

WATERVLIET ARSENAL

RESEARCH & ENGINEERING DIVISION

WATERVLIET NEW YORK

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Ti 6Al-6V-2Sn Alloy

Weld heat-affected zone

Charpy impact test

$\alpha \rightarrow \beta$ Transformation

X-ray diffraction

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A STUDY OF WELD HEAT AFFECTED-ZONES IN THE TITANIUM - 6Al-6V-2Sn ALLOY

Abstract

A new high-strength alpha-beta type titanium alloy was recently developed which is heat treatable to useful yield strengths above 180,000 psi, with 7 percent elongation, 16 percent reduction in area, and 7 ft. -lbs. Charpy V-notch impact energy at -40°F. Preliminary manual welding experience with this alloy disclosed a strong tendency for cracking in the heat-affected zone. This study was performed to determine resultant toughness in the heat-affected zones for various welding conditions. Preheat was the most influential factor in retaining toughness in solution treated and aged base metal; welding without it caused severe losses in impact strength and notched tensile strength which could not be recovered by aging treatments afterwards or resolution treating and aging. Mechanisms causing changes in properties involving $\beta \rightarrow \alpha$ and $\beta \rightarrow \alpha'$ on cooling, and $\alpha' \rightarrow \alpha + \beta$ and $\beta \rightarrow \alpha$ upon reheating were studies by X-ray diffraction volumetric analysis and photomicrographic techniques. No evidence of ω phase forming or transforming was found in the aging procedures studied.

Cross-Reference Data

Ti 6Al-6V-2Sn Alloy

Weld heat-affected zone

Charpy impact test

$\alpha \rightarrow \beta$ Transformation

X-ray diffraction

DO NOT REMOVE THIS ABSTRACT FROM THE REPORT

CONCLUSIONS

1. Toughness in the weld heat-affected zone of the Titanium - 6Al-6V-2Sn alloy is sharply decreased when manually welded without preheat.

2. Aging of manual welds in this alloy, made without preheat, does not help recover toughness in the weld heat-affected zone most critical area.

3. Preheating this alloy to 500°F permits a maximum retention of toughness in manual welds performed on heat-treated base metal.

4. Re-solution treating and aging a manual weld, made without preheat, does not help recover toughness in this alloy at all, but decreases toughness even more.

5. The technique of quantitative phase analysis of $\alpha + \alpha'$ and β by X-ray diffraction integrated intensities is most helpful in determining mechanisms of formation and decomposition of phases present.

6. Micrographic analysis of structures present in this alloy is only partially helpful for determination of phases present and possible toughness properties in the weld heat-affected zone.

7. The Rensselaer Polytechnic Institute synthetic specimen technique is readily adaptable to mechanical and physical metallurgy investigations of weld heat-affected zones in titanium alloys.

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INTRODUCTION

In this investigation the alloy studied was the Ti - 6Al-6V-2Sn (Fe, Cu) composition, developed by New York University for Watertown Arsenal Laboratories,[1] which is capable of reaching 180,000 psi (0.1 percent offset) minimum yield strength with 7 percent minimum elongation, 16 percent minimum reduction in area, and 7 ft. -lbs. minimum impact strength at -40°F. A lack of detailed knowledge of the behavior of this alloy, when welded, has prevented several fabricators from using the alloy. Lack of information on welding this alloy prevented its selection for prototype third stage Minuteman rocket motor cases.[2] Cracking of a Davy Crockett bracket made of this alloy was attributed by one investigator to be due to gaseous contamination in a tack weld being subsequently dissolved into the fusion zone by a weld pass.[3] Later another investigator showed the susceptibility of this alloy to weld thermal cycles,[4] wherein a specimen of Ti - 6Al-4V thermally cycled to 2400°F had a maximum free bend angle of 17 degrees, whereas the Ti - 6Al-6V-2Sn alloy broke at 2 degrees.

Military interest in high strength titanium is well established. Specifications [5-7], have continuously been upgraded to make use of the advantageous properties of titanium alloys and establish minimum standards for material supplied. The Ti - 6Al-6V-2Sn alloy is considered to have an important role in meeting specification requirements at the 170,000 - 180,000 and 180,000 - 190,000 psi (0.1 percent offset) yield strength levels in bar, billet, forgings, and extrusions. Meetings between Army producers and suppliers have emphasized potential usefulness of the alloy[8]. The interest in developing optimum welding procedures therefore follows.

The Rensselaer Polytechnic Institute synthetic technique for producting weld heat-affected zone test specimens was chosen.[9,10] Application of this technique to weld heat-affected zone studies has been demonstrated for the 6Al - 4V[10], Ti - 150 A[11], 4 Cr - 2Mo and the 7 percent Mn[12] alloys of titanium. Existing weld thermal cycle data for 1/4 -inch Ti - A55 plate was obtained for use in this study.[13] Reference 11 concluded that thermal data was applicable to low alloys of titanium as well as to pure titanium which obviated recording of weld thermal cycles in the 6Al - 6V-2Sn alloy. Properties studied were Charpy V-notch impact strength, Vickers (30Kg.) hardness, and notched ($K_t = 5.5$) tensile strength. Photomicrographs at 260 X were obtained as were volume amounts of alpha plus alpha-prime and beta phases in the specimens. The latter results were obtained by X-ray diffraction technique.

OBJECTIVE

The purpose of this investigation was to determine the properties in the weld heat-affected zones of a very high strength alpha-beta type titanium alloy and determine the effect of weld variables and heat-treatment on these properties.

MATERIAL

The alloy studied in this investigation was prepared by Titanium Metals Corporation of America. Starting bar stock 2 1/4-inches square was hot rolled square to oval to square, etc., reducing 13 to 17 percent per rolling pass. Temperature of the bar at the beginning of rolling was 1700°F and 1400 to 1600°F at the finish. 140 feet each of 1/4-inch round and 1/2-inch square bar was produced this way. Chemical analysis was as follows:

Element	Weight Percent	
	Heat No. M-9520	Required
Al	5.6	5.2 - 5.8
V	5.4	5.2 - 5.8
Sn	2.0	1.7 - 2.3
Fe	0.51	0.5 - 0.9
Cu	0.28	0.2 - 0.5
C	0.026	0.05 max.
O_2	0.1785	0.10 - 0.19
H_2	0.008	0.01 max.
N_2	0.016	0.01 - 0.035

After the bars were rolled, they were pickled, then solution treated at 1650°F for 1/2 hour, water quenched, aged at 1000°F for 4 hours, and air cooled. The following resultant mechanical properties were determined by Titanium Metals Corporation of America:

Property	Value		
	Min. Specified *	Measured	
		1/2-Inch Square	1/4-Inch Round
Yield Strength (.1 percent offset)	170,000 psi	188,000	171,000
Yield Strength (.2 percent offset)	-	193,000	178,000
Ultimate Tensile Strength	-	205,000	183,000
Elongation in 2-inches	7 percent	7	13
Reduction of Area	14 percent	15-20	40
V-notch Charpy impact at -40°F	7 ft.lbs.	12	-

* Military Specification, WA-PD-76C(1), 12 March 1956

Charpy V-notch impact tests were also performed for the as-received (Solution treated and aged) 1/2-inch square bar after receipt from the supplier and disclosed a value of 9.3 ft.lbs. at -40°F, average of three specimens. This impact strength was used in all the graphs. It should also be noted that the Mil. Spec.[5] used for acquisition of this material was superseded by new Mil. Specifications.[6,7]

SPECIMEN PREPARATION

Impact Specimens

The 1/2-inch square bar stock, as received, was cut into specimen blanks 0.420-inch square and 2.80-inch long. Thermocouples were then percussion welded at the mid-point, after which thermal cycling in a time-temperature controller was performed. Standard V-notch Charpy bars were machined from each thermally cycled blank, locating the notch in the center of the peak temperature zone produced at the mid-section. A uniform amount of material was removed from each side of the blank to produce the .394-inch dimensions. See figure 1.

Notched Tensile Specimens

The 1/4-inch round bar stock, as received, was cut into lengths of 4.0-inches long. Thermocouples were affixed at mid-point and the specimen blanks were then thermally cycled to produce the desired microstructure for select points in the weld heat-affected zone. After thermal cycling, notched tensile bars were machined as shown in figure 2. The geometry for the notch was determined from H. Neuber's data[14] and for an assumed stress concentration factor of 5.5. It did not appear feasible to go to higher values of stress

concentration for this size bar and minor diameter without endangering test results reliability.

After thermal cycling, some specimens were heat-treated in a controlled atmosphere furnace, and then finish machined as above.

Metallographic Specimens

Specimens for X-ray diffraction analysis, photomicrography, and hardness measurements were prepared from as-received and thermally cycled 1/2-inch square bars by sectioning each specimen at right angles to the longitudinal axis. The thermally cycled specimens were sectioned 1/32-inch past the thermocouple (peak temperature) control point. Mounting of the sections was in lucite. Each mounted specimen was then used for appropriate X-ray, photomicroscopy, and hardness measurements after 1/64 to 1/32-inch was removed by a wet sanding wheel and appropriately polished or etched.

METHOD OF PROCEDURE

Impact Test

The microstructures at select points in the weld heat-affected zone were duplicated in the center of the 1/2-inch square specimen blanks using an improved version of the time-temperature controller developed at Rensselaer Polytechnic Institute.[9] This apparatus resistance heats the specimen clamped between water cooled copper jaws; the specimen, plus jaws, forming the secondary loop of a welding transformer. Electronic control of the time-temperature cycle was obtained by a thermal history programmer and a feed-back temperature sensing circuit using a thermocouple percussion welded to the outside surface mid-point of the specimen which was in turn centered between the clamping jaws. Platinum - platinum/10 percent rhodium thermocouple wires, .010-inch diameter each, were used for maximum dependability to above 2400°F.[15] It was found that the two wires had to form a junction with the specimen surface, such that, the plans established by the two wires leading immediately into the junction had to be at right angles to the longitudinal axis of the specimen. Deviation from this caused transient voltages to be induced in the thermocouple circuit other than from thermoelectric effects, causing errors to be produced in the temperature sensing a and control circuit. Proper positioning of the thermocouple permitted maintenance within $\pm$ 5°F of the programmed temperature-time cycle.

An optimum clamping distance of 0.80-inch between the copper jaws was found. Larger clamping distances prevented the mid-point along the specimen from cooling as fast as programmed, due to the limiting rate of heat conduction in the bar from this mid-point to the copper jaws. Smaller clamping distances other than optimum, progressively narrowed the width of the uniformly heated mid-range along the specimen, thus making sectioning and location of notch at mid-point all the more critical.

Impact tests were performed by the Watertown Arsenal Materials Properties and Testing Laboratory using a Mouton 217 ft.-lb. impact testing machine.

A series of impact specimens were thus produced for synthetic weld thermal cycles having peak temperatures of 1400, 1600, 1800, 2000, 2200, and 2400°F representing 25,000 joules per inch energy input (typical for manual welding of 1/4-inch titanium plate at 5 inches per minute welding speed) and based on existing F(s,d) data and 80°F initial plate temperature.[13] See figure 3 for typical weld thermal cycles used.

A series of impact specimens were also thermally cycled to 2400°F peak temperature and subsequently aged at 900, 1000, 1100, and 1200°F, each for 1/2, 2, and 4 hours aging time. Aging was done in an electric furnace, argon filled, and preheated to the desired temperature. Specimens were permitted to heat on the fire-brick hearth. After aging for the desired time, the specimens were removed from the furnace and cooled in argon to room temperature.

The synthetic weld thermal cycle was also determined and specimens produced for 2400°F peak temperature, 25,000 joules per inch energy input, and 500°F initial plate temperature (preheat). Some specimen blanks experiencing 2400°F peak temperature for the weld thermal cycle based on 25,000 joules per inch and 80°F initial plate temperature were subsequently solution treated at 1650°F for 2 hours, water quenched, then aged at 1100°F for 4 hours and air cooled.

Notched Tension Test

Round specimen blanks, 1/4-inch diameter, were thermal cycled in the same manner as the impact specimens. As-cycled specimens were produced for 1400, 1600, 1800, 2000, 2200, and 2400°F peak temperatures. Aging treatments of 900, 1000, 1100, and 1200°F for 1/2, 2, and 4 hours were investigated. Specimens for the 500°F preheat, 2400°F peak temperature cycle were produced, as were re-solution treated and aged specimens. The notched tensile bars were then tested in a Riehle tensile testing machine, model FS-60, using flat wedge grips and a crosshead velocity of .005 inches per minute. Load at failure was recorded. There was no measurable reduction in area for these specimens.

Microstructure and Hardness Measurement

Specimens for microstructural analysis were polished on a 600 grit wet wheel, followed by three lapping wheels using 20 to 1 μ diamond paste progressively, and then followed by 0.3 μ Alundum on the last lapping wheel. The surface was etched lightly with Margolin's "R" etchant[17] and then repolished on the lapping wheels. The "R" etchant composition used was:

13 ml - 10 percent Benzalkonium Chloride
35 ml - Ethyl Alcohol
40 ml - Glycerine
10 ml - 50 percent Hydrofluoric Acid

and was found to be much easier to control and avoid excessive grain boundary

attack and scattered pitting as compared to a variety of the HF - HNO_3 base etchants. After repolishing, the surface was etched again by swabbing once every 3 to 8 seconds. As this alloy studied was high in alpha and alpha-prime, etching was repeated until the alpha-prime became well developed and alpha phase started to delineate. A standard metallograph was then used to make photomicrographs at a magnification of 260 diameters.

Hardness was determined for each specimen using a Vickers hardness tester at 30 kg. load and a reflex-type direct measuring optical system. An average of five readings was made to determine the reported hardness.

X-Ray Analysis

Metallographic methods of phase analysis of titanium alloys generally are subject to much error. The Ti - 6Al-6V-2Sn alloy was especially difficult to define metallographically due to the low (10 to 30) percentage of beta phase present. Point counting and linear analysis was not applicable accurately because of the difficulty of resolving the beta microconstituents metallographically. Polarized light techniques had been applied to attempt to identify the anisotropic alpha phase from the isotropic beta phase, but alpha exhibited changing birefringence due to orientation. This meant that only portions of alpha refracted at any one rotational angle, never 100 percent. Also, the alpha crystal would exhibit no birefringence when its c-axis was parallel to the surface illuminated.

Averbach and Cohen had defined the application of X-ray diffraction procedures for retained austenite quantitative analysis.[18] Later studies by Lopata[19] and Averbach, et al.,[20] disclosed the applicability of the same technique to quantitative phase analysis of titanium and its alloys and accuracies and errors associated with the technique. Lopata's technique was rather closely followed in this study with some minor improvements. A slower goniometer sweep rate of 1/8 degree per second was used. A standard 50 kv Norelco diffractometer with goniometer attachment was operated at 45 kv and 18 ma. A copper target provided CuK_α radiation. The CuK_β radiation and titanium fluorescence was reduced by a diffractometer discrimination attachment. A one-degree divergence, .006-inch wide receiving, and no scatter slit were used.

The recorder scale was set to have the most intense peak, $(10\bar{1}1)_\alpha$ employ the full height of the in-line strip chart, plotting intensity vs. angle directly. The first three diffraction planes of alpha phase and the first diffraction plane of beta phase were studied. These occurred over the range of 34 to 42° 2 θ Bragg Angle. The $(10\bar{1}1)_\alpha$ and $(110)_\beta$ diffraction peaks were specifically selected from prior experience based on the large intensities obtained and the low error involved in the presence of moderate amounts of preferred orientation. The integrated intensities of the peaks selected were obtained by measuring areas under each peak with a compensating polar planimeter. These areas were substituted into the equation:

$$\frac{K\,a}{b} = \frac{f_\alpha}{1 - f_\alpha}$$

where: $K = 1.41$ (the ratio of the integrated intensity of the $(110)_\beta$ peak of a pure beta specimen to the integrated intensity of the $(10\bar{1}1)_\alpha$ peak of a pure alpha specimen); a = the area measured under the $(10\bar{1}1)_\alpha$ peak; b = the area measured under the $(110)_\beta$ peak, and f_α is the volume fraction of alpha (plus alpha-prime) phase present. This can be further simplified to the form:

$$f\alpha = \frac{X}{1 + X}$$

where $X = \frac{Ka}{b}$, terms previously defined.

Errors of less than 3 percent are possible, with this technique, if suitable care is taken in specimen preparation, curve fitting, and area plotting. Both mechanical polish and electropolish techniques were compared and for the procedures used with a difference of less than 3 percent. The peaks selected for study overlapped and therefore individual areas for each peak had to be obtained by extrapolating the $(110)_\beta$ peak to background, assuming symmetry of this peak existed. Some skewness of the peak $(10\bar{1}1)_\alpha$ was observed, probably due to a slight difference in lattice parameters between alpha and alpha-prime (martensite) caused by differences in alloy content. This prevented using the $(10\bar{1}1)_\alpha$ peak symmetrical extrapolation for area separation from the $(110)_\beta$ peak as the degree of skewness did vary.

The effect of preferred orientation was minimized by selecting the $(10\bar{1}1)_\alpha$ plane for computation. Lopata studied preferred orientation in the Ti - 6Al-4V alloy by obtaining X-ray patterns from six sides of a pentagonal specimen, and found the pyramidal plane reflection varied only slightly and was less than the basal or prismatic plane reflection variations.[19]

Other errors associated with primary and secondary extinction and change in the value of atomic scattering factor with alloying, although present, were determined negligible.

RESULTS AND DISCUSSION

Effect of Peak Temperature on Properties

Impact strength dropped off rapidly as peak temperature of the thermal cycle increased. Figure 4 summarizes this effect. A volume of the base material in the heat-affected zone reaching peak temperatures above 1600° will have lower impact properties than the base metal. This low impact zone has a depth of about .080-inch from the fusion zone - heat-affected zone interface when welding is performed in one pass with 25,000 joules per inch energy input and the 1/4-inch thick plate is initially at 80°F. The 2200 - 2400°F peak temperature zone is most important which is just adjacent to the fusion zone and has an impact strength of only 5 1/2 to 6 ft. lbs. at -40°F. Results over the whole testing temperature range of -40°F to +80°F, shown in figure 5, reveal poor impact strength at any temperature for 2200 and 2400°F of peak temperature thermal cycles.

The 1400°F peak temperature thermal cycle raised the as-received impact strength somewhat, apparently due to over-aging.

Figure 6 shows the results of notched tensile tests. Notched tensile strength after the 1400°F thermal cycle was lower than before cycling, and with increase in peak temperature, the notched tensile strength progressively decreased more, being only 70 percent of the as-received strength for the 2400°F cycle. Thus, when the plate was welded without preheat, a large amount of the notch tensile strength was lost in the entire heat-affected zone.

Hardness was lower for thermal cycle peak temperatures up to 1800°F, and above that temperature the hardness increased, being about 10 percent higher at 2400°F. Figure 7 summarizes these data.

Thus properties are generally affected one way below and another way above a peak thermal cycle of 1800°F. The exception was notched tensile strength which was always lower after thermal cycling than before, regardless of peak temperature.

Effect of Peak Temperature on Microstructure

Figures 8, 9, and 10 show microstructures before thermal cycling and after each peak temperature thermal cycle. It appears that very fine alpha-prime existant in the as-received condition (solution treated at 1650°F for 1/2 hour, water quenched, aged at 1000°F for 4 hours, air-cooled) is being transformed into very fine alpha and beta on heating to 1400 and 1600°F. The large alpha platelets appear to get smaller and the small alpha platelets disappear upon heating to 1400°F and higher. This would be a diffusion type transformation of alpha to beta.

The equilibrium transformation diagram is approximated by figure 11 from Rausch's data[21] and one would expect to elevate the beta transus upon heating rapidly and suppress it upon rapid cooling. Cooling from 1400 and 1600°F peak temperatures did not appear to be fast enough to form acicular alpha-prime, at least any that was resolvable. However, some of the beta would be expected to transform by nucleation and growth to alpha. The overall result would be that there was less alpha plus alpha-prime after heating and cooling than before. The X-ray diffraction studies confirmed this, as seen in figure 12. This figure is not capable of indicating what amount of alpha-prime transformed to alpha for these thermal cycles, but figure 9a and b encourage the view that a significant amount did.

Thus, some alpha-prime transformed to alpha, and some alpha transformed to beta so that the total amount of beta increased for the 1400 and 1600°F peak thermal cycles. This explains the observed increase in impact strength, decrease in notched tensile strength, and decrease in hardness.

Upon heating to 1800°F, the same changes occurred, but upon cooling some of the beta was transforming to alpha-prime as well as very fine alpha, due to the relatively high cooling rate. Figure 12 shows that hardly any

change in the total amount of beta took place as a result of the thermal cycle. Hardness didn't change appreciably, nor did impact energy at -40°F. Above testing temperatures of -10°F, an improvement in impact strength occurred. Notched tensile strength was still lower after cycling, but higher than for the 1600°F thermal cycle.

Heating to 2000, 2200, and 2400°F by thermal cycling certainly betanized most, if not all, the alpha-prime and alpha existing before. As peak temperature went up, the cooling rate also increased, thus forming more alpha-prime on cooling as well as alpha. The overall effect was to have more grain growth occur during the cycle, less beta retained, and more alpha-prime form upon cooling. Thus, impact strength continued to drop off, notched tensile strength dropped more, and Vickers hardness increased more as 2400°F peak temperature was approached in the thermal cycle.

Effect of Aging on Thermal Cycled Properties and Microstructure

As mentioned before, specimens for aging were first thermally cycled to 2400°F peak temperature representing welding conditions of 25,000 joules per inch energy input and 80°F initial plate temperature. Isothermal transformation characteristics for this alloy were not established.

Effect of aging on properties in Ti - 6Al-6V-2Sn welds has only been partially investigated by others[22,23] and is not complete. Effect of aging on properties in other alpha-beta titanium alloys has been studied by many. Pronounced age hardening occurring in very short times was a puzzle to many early investigators[24 - 28] who observed it.

This later proved to be the omega phase unresolvable by metallographic techniques but observable[29] by select X-ray diffraction techniques. However, figure 13, hardness vs. aging time shows no sudden change within the first hour that indicates presence of ω. In fact, over long aging times there are no strong trends except that the higher peak temperatures do cause more of a decrease in hardness with time. Figure 14 helps to explain this. Upon heating to any aging temperature below the Beta transis and holding isothermally, two reactions are competing: one is $\alpha' \rightarrow \alpha + \beta$, and the other is $\beta \rightarrow \alpha$, both of which are diffusion reactions. From 0 to 1/2 hour at 900 and 1000°F, the $\beta \rightarrow \alpha$ reaction appears to dominate, and after 1/2 hour the $\alpha' \rightarrow \alpha + \beta$ reaction appears to dominate. What makes the hardness data appear insensitive is that one reaction increases hardness and the other decreases hardness, so the apparent hardness may change very little if both reactions do occur.

On heating from room temperature to 1100°F, the balance between the two above reactions was such that by 1/2 hour there was hardly any change in beta phase volume. After 1/2 hour, beta volume increased as the $\alpha' \rightarrow \alpha + \beta$ reaction dominated. On heating from room temperature to 1200°F, the volume percent of $\alpha + \alpha'$ decreased from 85.3 percent as the $\alpha' \rightarrow \alpha + \beta$ reaction dominated. Then, after 1/2 hour, the reaction, $\beta \rightarrow \alpha$, was forming more total alpha phase. Similar behavior in other alloys had also been observed.[30-37]

Impact energy generally suffered as a result of any aging time or temperature. Figure 15 shows this behavior. The reason for this is not well substantiated, even by the microstructures shown in figures 16, 17, 18, and 19. Whatever it is that happened, it occurred within the first half hour of aging.

The notched tensile data are shown in figure 20 and also are not readily explainable. Notched tensile strength increased within the first 1/2 hour of aging from 161,000 psi to 176,000 - 197,000 psi. After four hours the same range of notched strength prevailed, although minor trends were noted.

Effect of Preheat on Properties and Microstructure

The specimens thermally cycled to 2400°F for 25,000 joules per inch energy input and 500°F initial plate temperature disclosed remarkable improvements over non-preheated specimens as expressed by impact strength, and notched tensile strength. The microstructure is shown in figure 21a and discloses much more beta present than before, even though alpha-prime predominates. Figure 12 shows 32.2 percent beta present after cycling, instead of 14.8 percent. This is more than a 100 percent increase in beta retained upon cooling. The hardness, figure 7, is far below the unpreheated specimen at 2400°F; notched tensile strength is much improved, figure 6, over the non-preheated specimen. Impact strength is even better than the 2000°F peak temperature specimen without preheat, and rises very fast at testing temperatures above 20°F. Retention of the larger amount of beta upon cooling is undoubtedly due to the longer time at high temperature, and slower cooling rates. This can be seen in figure 3.

The data for this effect of preheat are certainly the most significant optimistic trends in this entire study.

Effect of Re-Solution and Aging Treatment on Properties and Microstructure

Re-solution treatment and aging of the specimen cycled to 2400°F peak temperature did not change the impact strength and notched tensile strength decreased by about 7 percent. Vickers hardness increased by about 5 percent and amount of $\alpha + \alpha'$ decreased by about 1.3 percent. The microstructure, as shown in figure 21, is very fine alpha comprising 84 percent by volume in a beta matrix comprising 16 percent by volume. The results discourage this treatment entirely as a technique to recover properties in as-welded structures.

SUMMARY

Welding of all of the ultra-high strength titanium alloys by conventional techniques can detrimentally influence the properties in the heat-affected zone. The routine problem is to determine how "best" to control the welding variables to optimize retention of strength and not lose toughness of the base metal.

A clear understanding of the kinetics and mechanisms of transformations involved in the weld heat-affected zone will most directly help investigators know: (a) why certain changes took place, and (b) how best to control the undesired changes from taking place when welding very high strength material. Techniques used in this investigation can also help others solve weldability problems in many other alloys, concisely and directly.

The encouragement gained in this study by using high preheat, certainly indicates further study of the effect of this preheat on properties in the rest of the heat-affected zone. Other preheats might also be studied for the same reason.

The time interval from 0 to 30 minutes for aging at 900 to 1200°F, if studied in increments, would disclose actual rates of transformations occurring in very short times in the heat-affected zone structures.

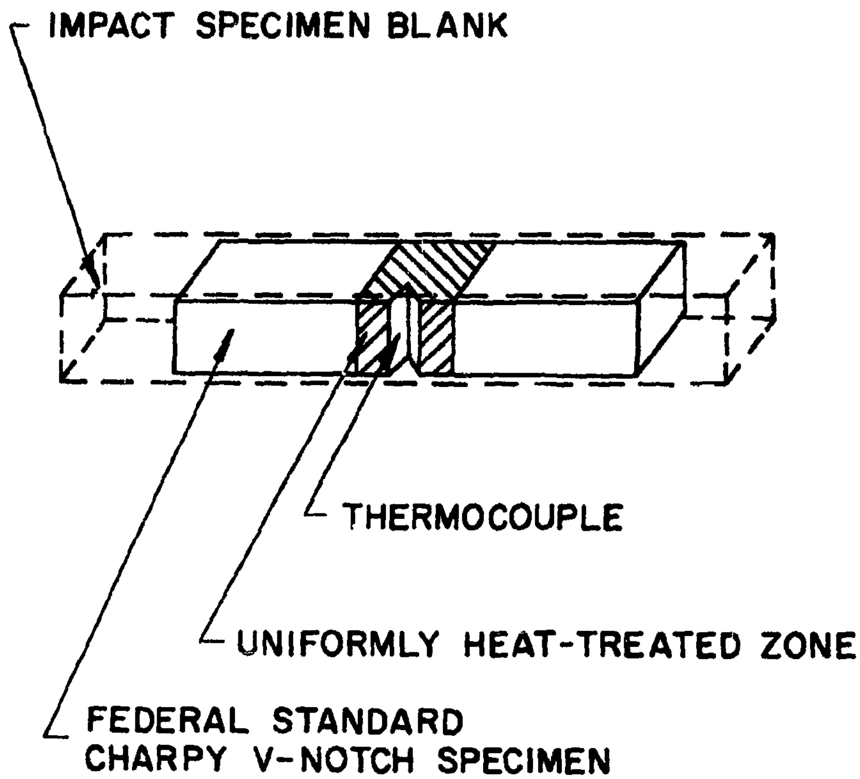


FIGURE 1. CHARPY V-NOTCH IMPACT SPECIMEN

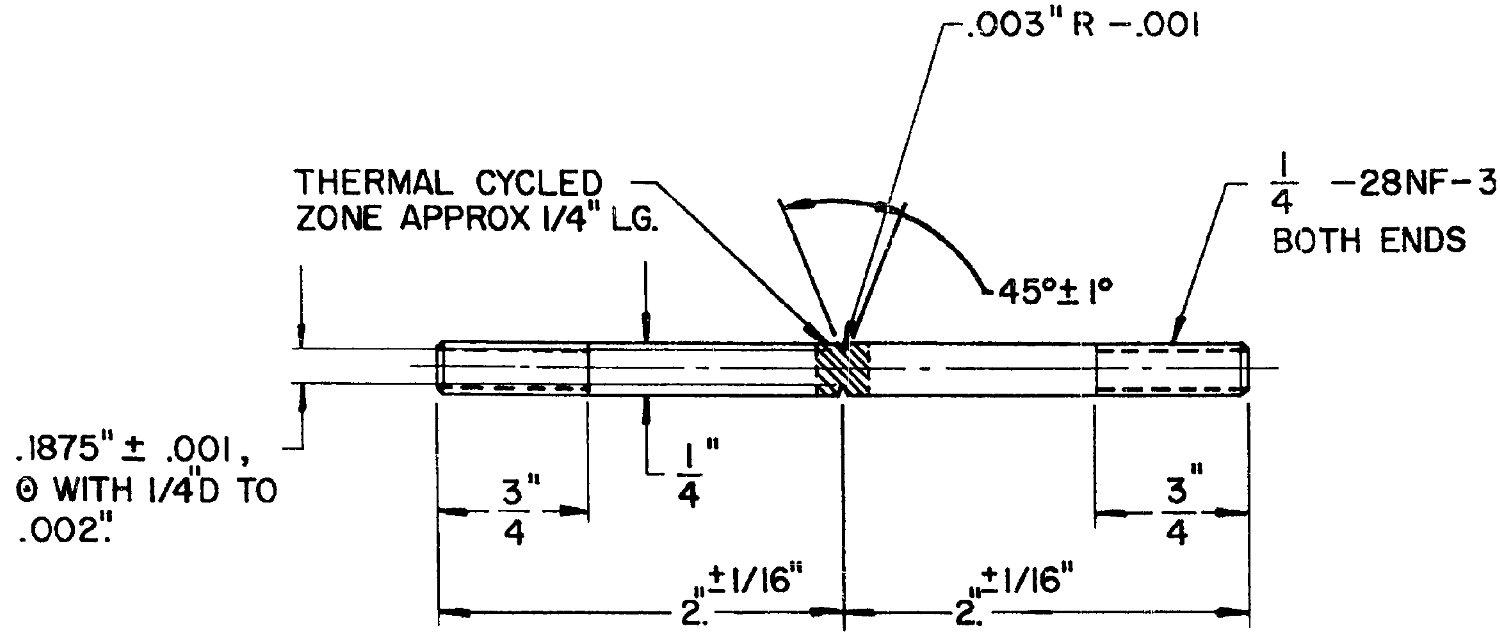


FIGURE 2. NOTCHED TENSILE TEST SPECIMEN $K_t = 5.5$ (H. Neuber)

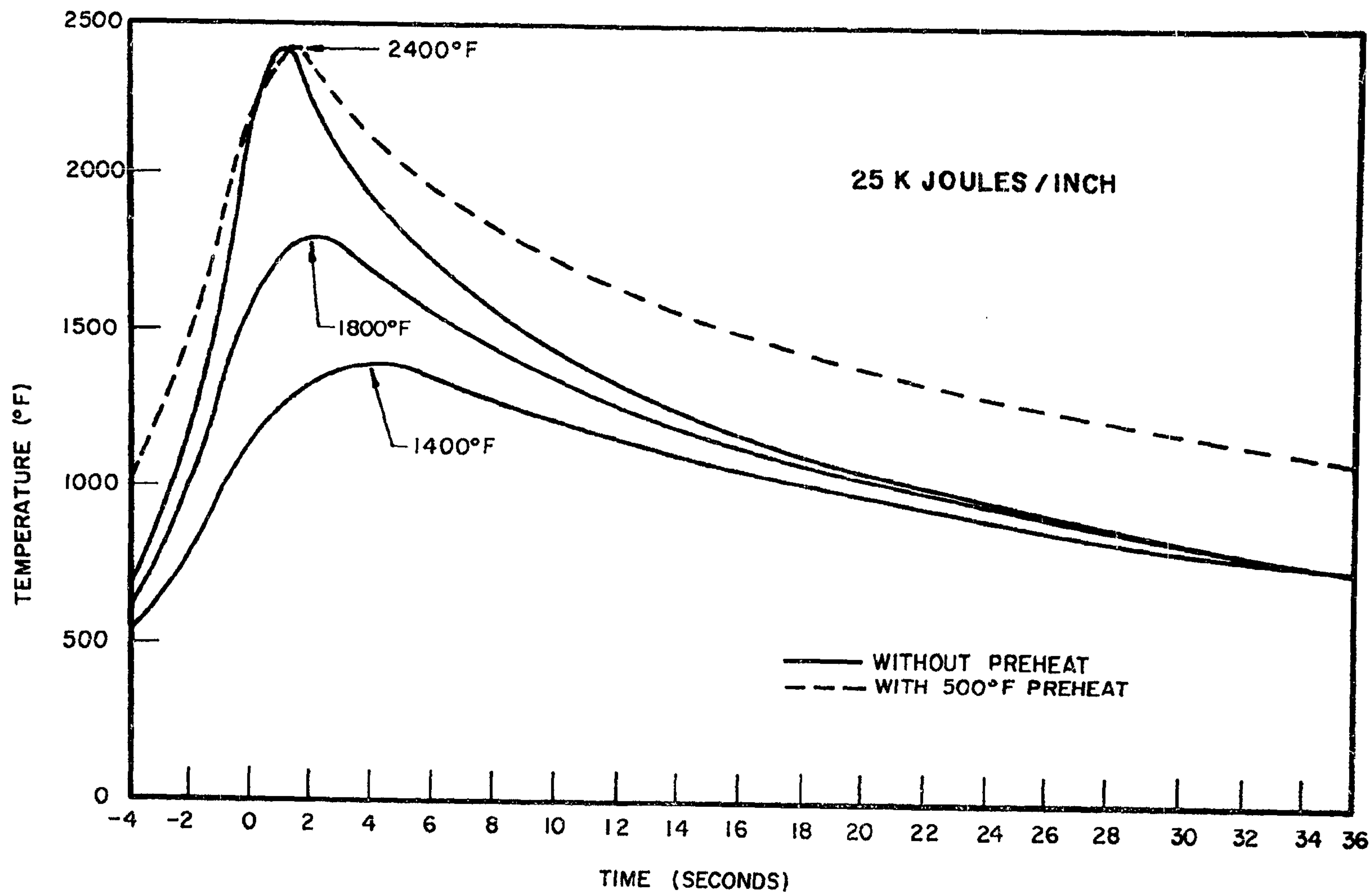


FIGURE 3. THERMAL CYCLES IN THE WELD HEAT-AFFECTED ZONE AT THREE DIFFERENT DISTANCES FROM THE WELD CENTER LINE IN 1/4 INCH THICK TITANIUM PLATE

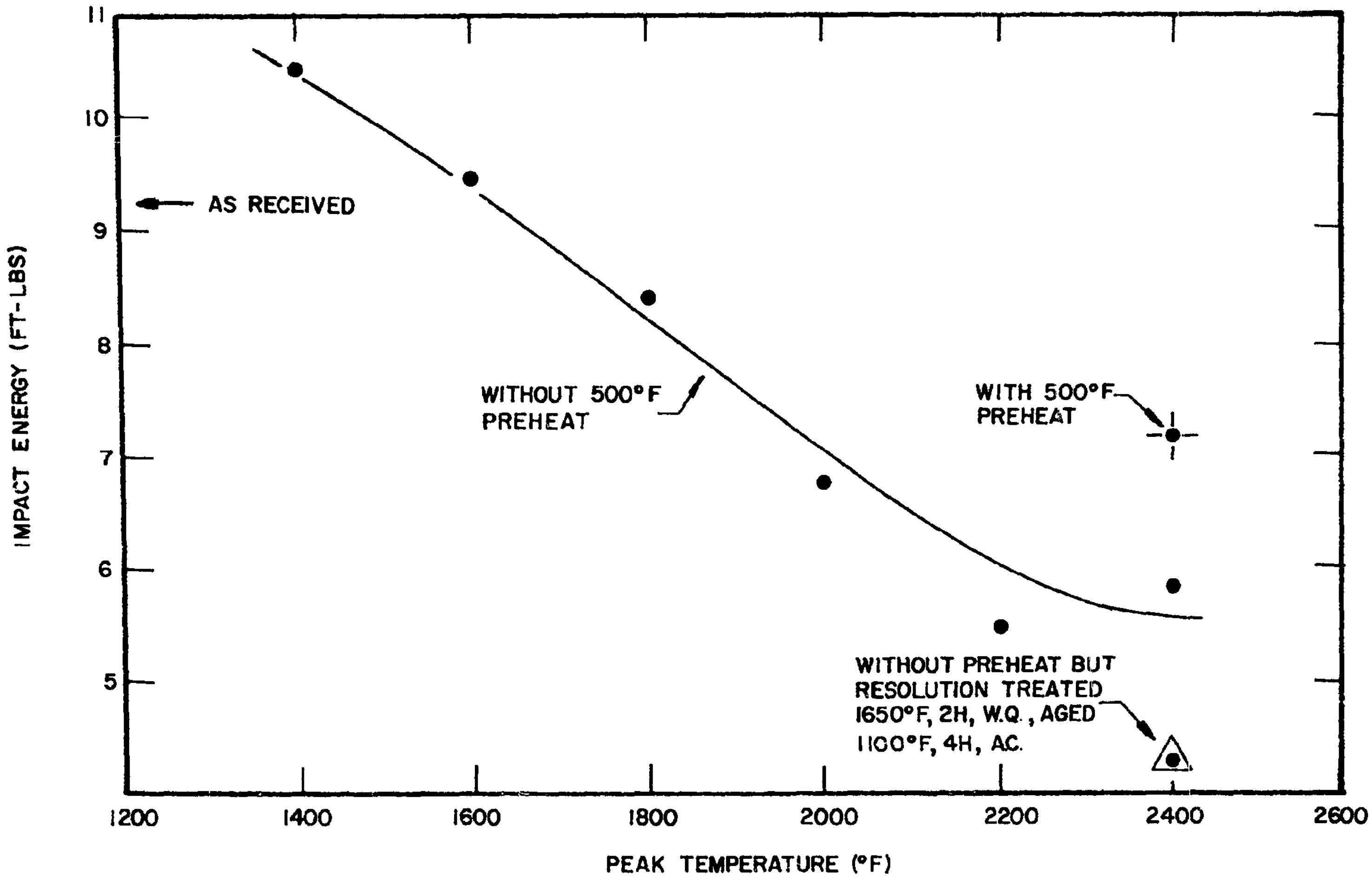


FIGURE 4. EFFECT OF THERMAL CYCLE PEAK TEMPERATURE ON CHARPY V-NOTCH IMPACT STRENGTH AT -40°F

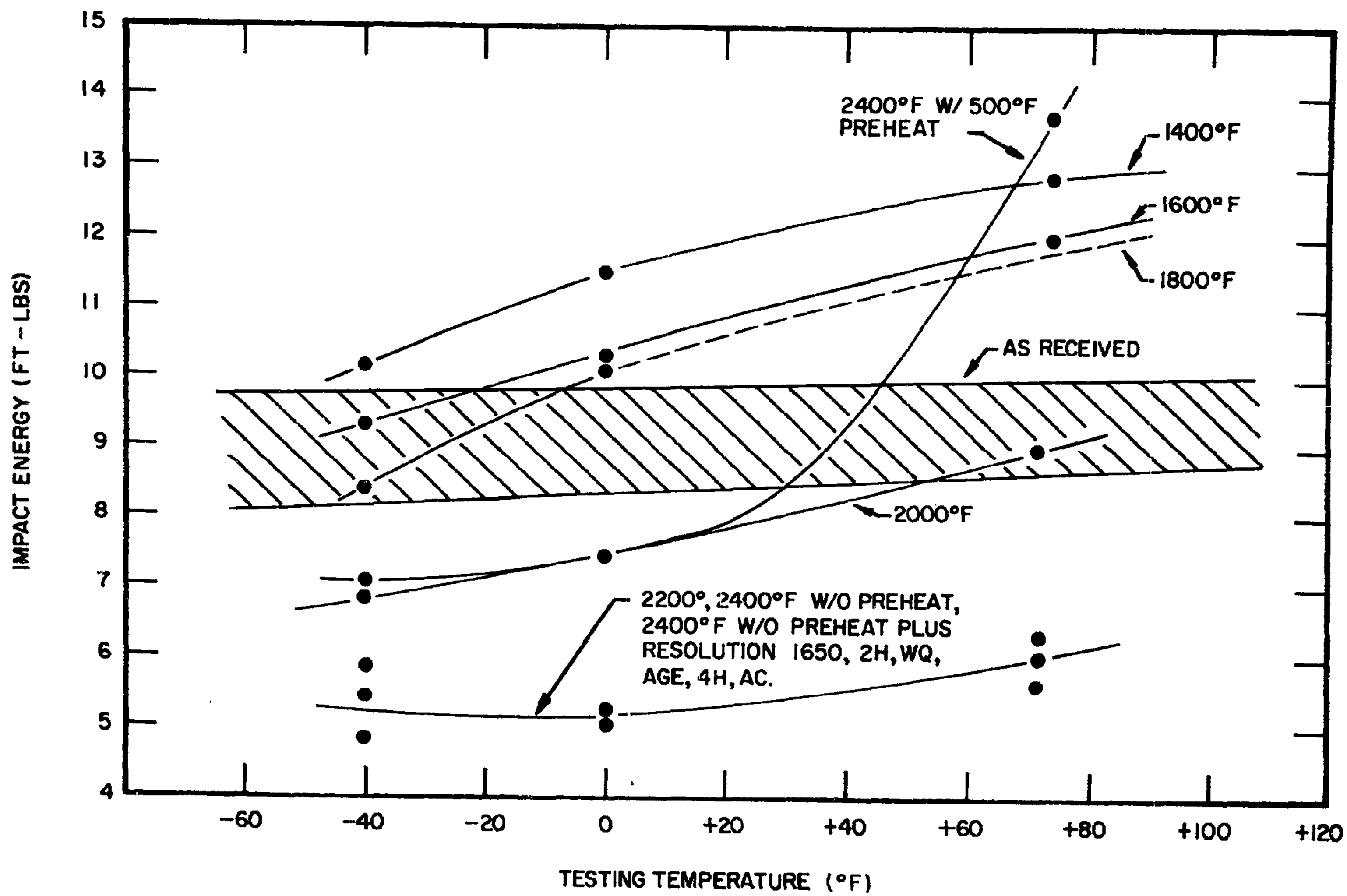


FIGURE 5. EFFECT OF TESTING TEMPERATURE ON CHARPY V-NOTCH IMPACT STRENGTH FOR VARIOUS THERMAL CYCLE PEAK TEMPERATURES

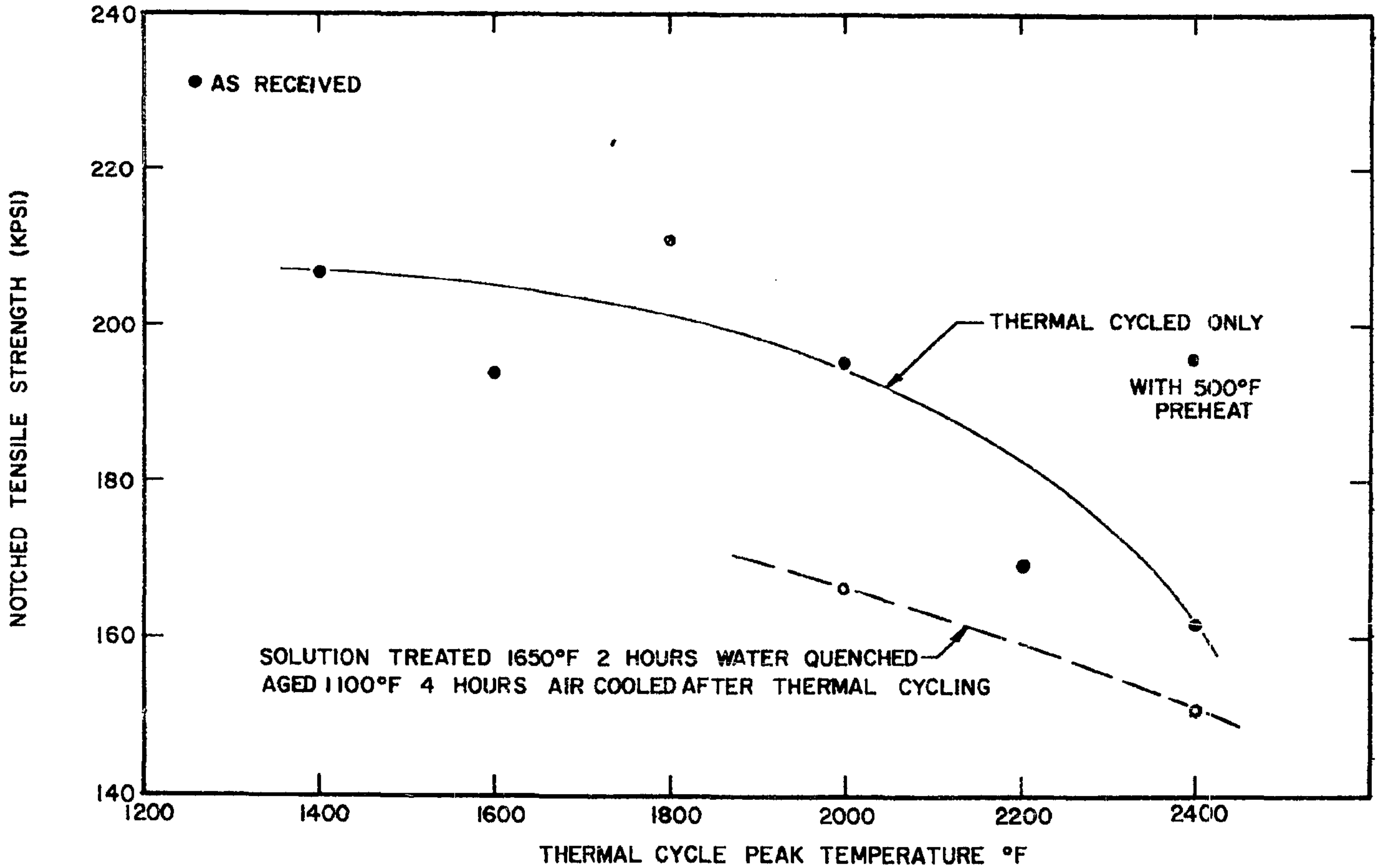


FIGURE 6. EFFECT OF THERMAL CYCLE PEAK TEMPERATURE ON NOTCHED TENSILE STRENGTH THERMAL CYCLED WITHOUT POST-HEAT

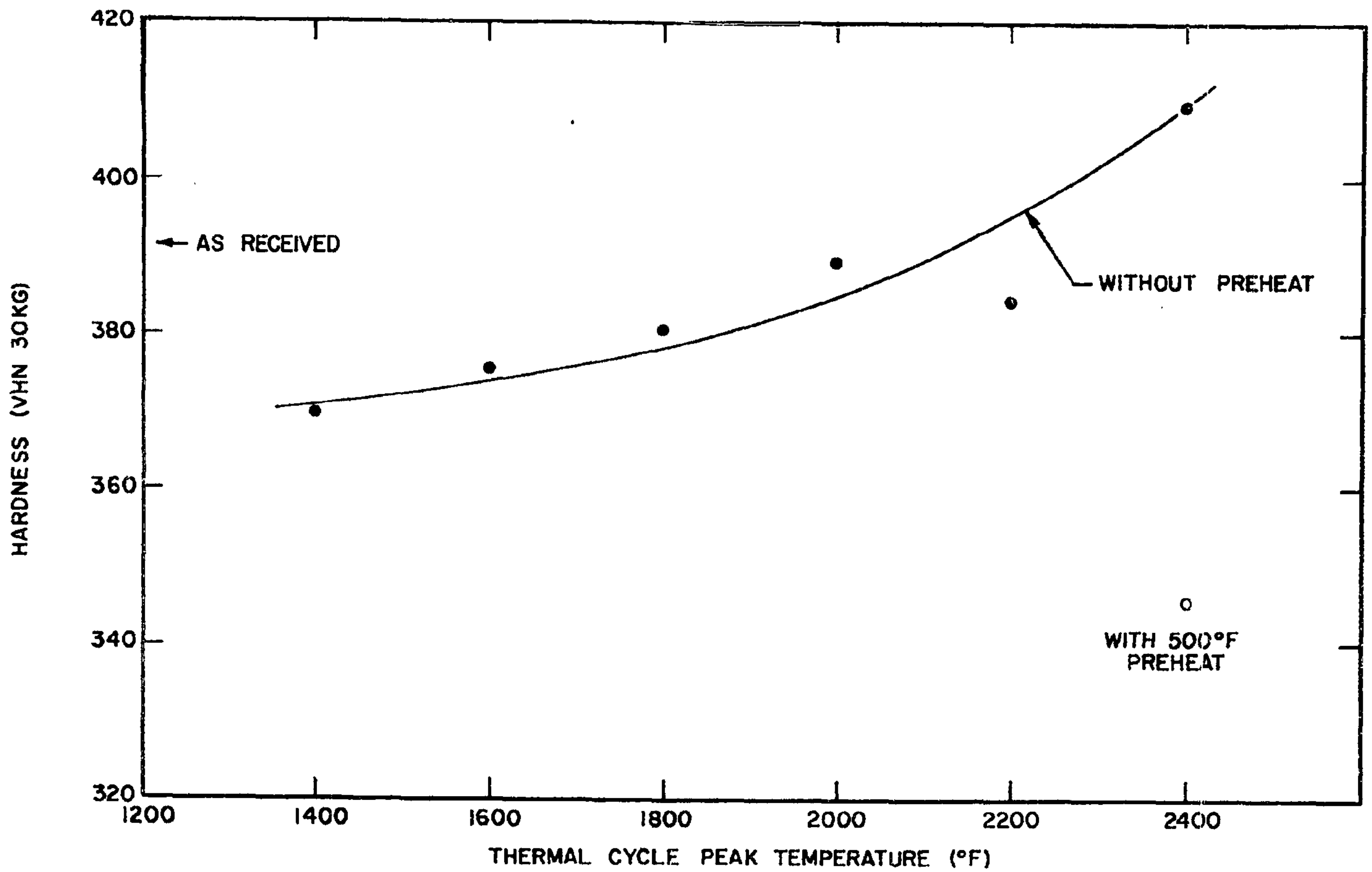


FIGURE 7. EFFECT OF THERMAL CYCLE PEAK TEMPERATURE ON HARDNESS (AS THERMAL CYCLED)

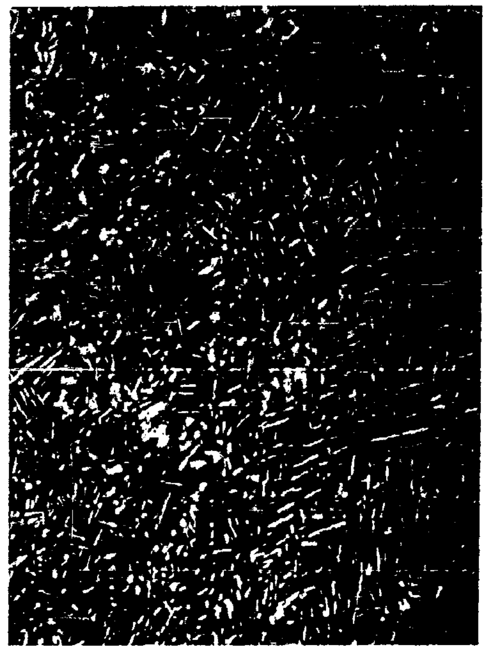
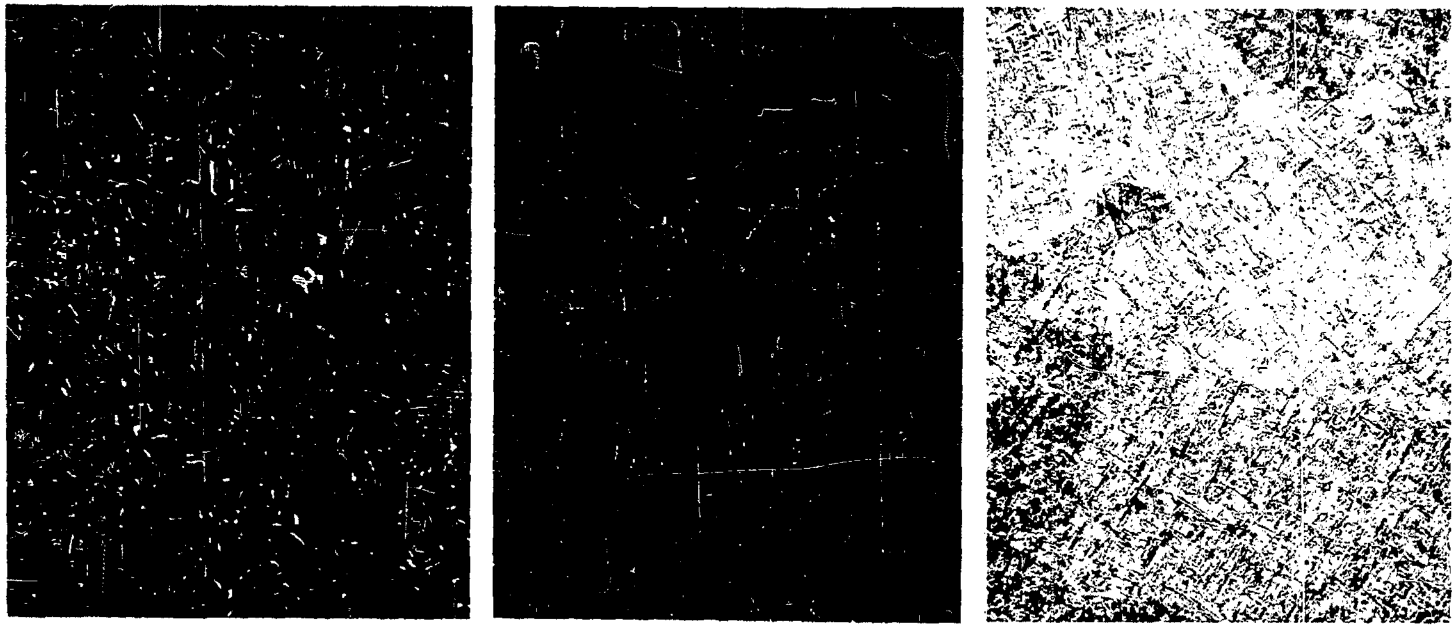
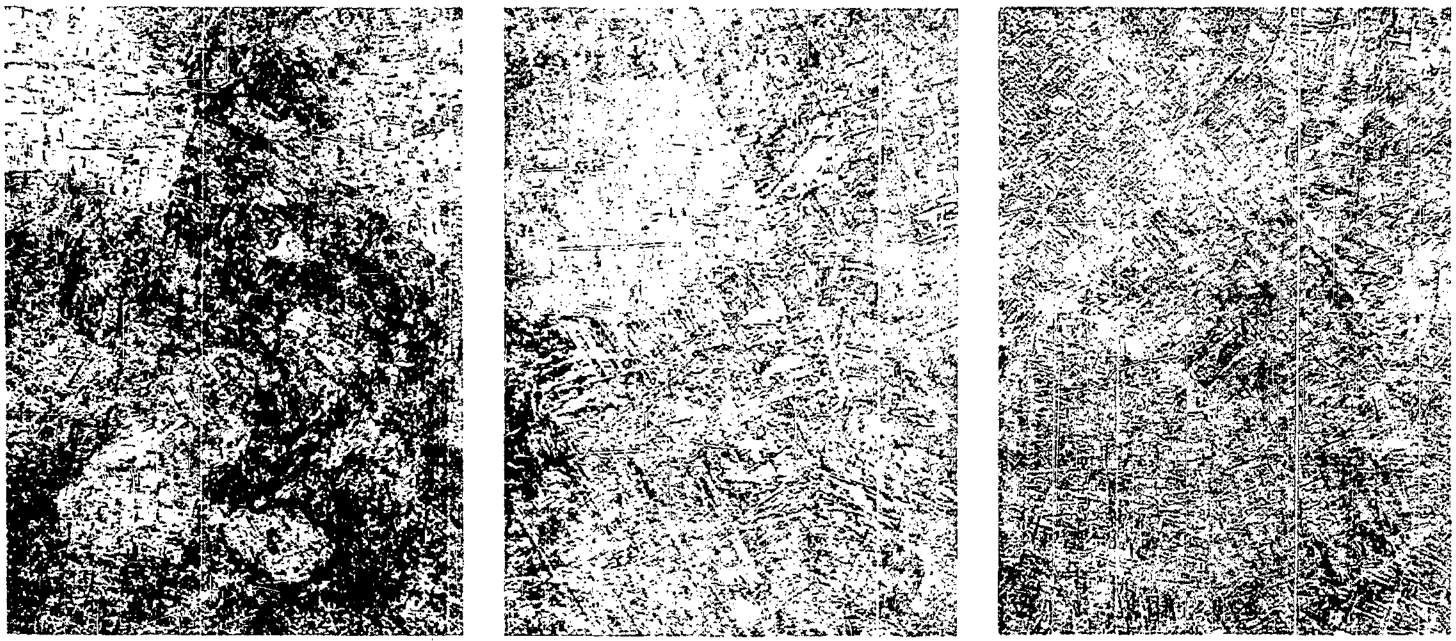


FIGURE 8. AS-RECEIVED MICROSTRUCTURE
260X 'R' ETCHANT



(a) 1400°F (b) 1600°F (c) 1800°F

FIGURE 9. AS THERMAL CYCLED TO PEAK TEMPERATURES INDICATED
260X 'R' ETCHANT



(a) 2000°F (b) 2200°F (c) 2400°F

FIGURE 10. AS THERMAL CYCLED TO PEAK TEMPERATURES INDICATED
260X 'R' ETCHANT

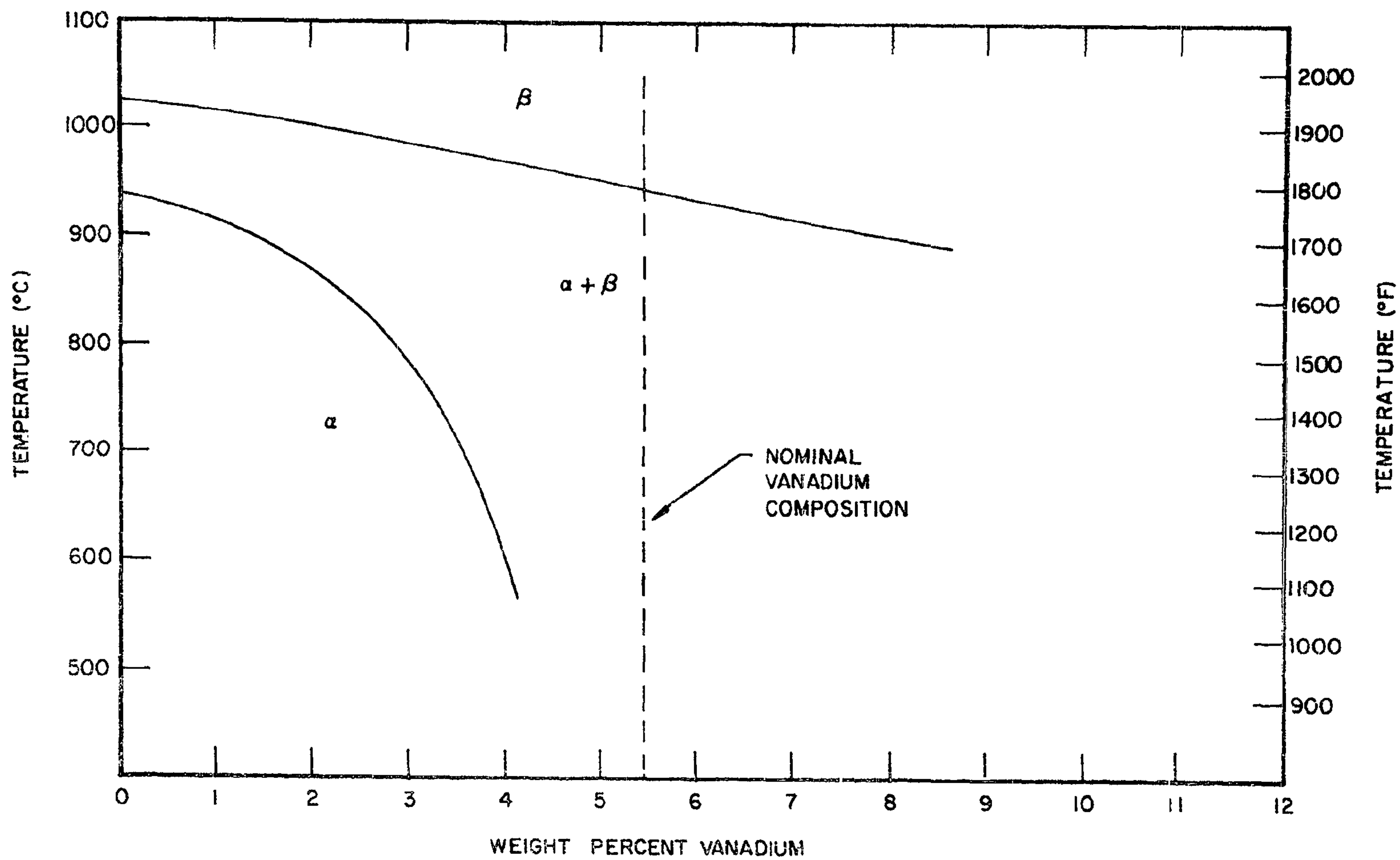


FIGURE 11. VERTICAL SECTION OF TI-AL-V SYSTEM AT 5.5 PCT. AL. BASED ON DATA BY RAUSCH, CROSSLEY, AND KESSLER

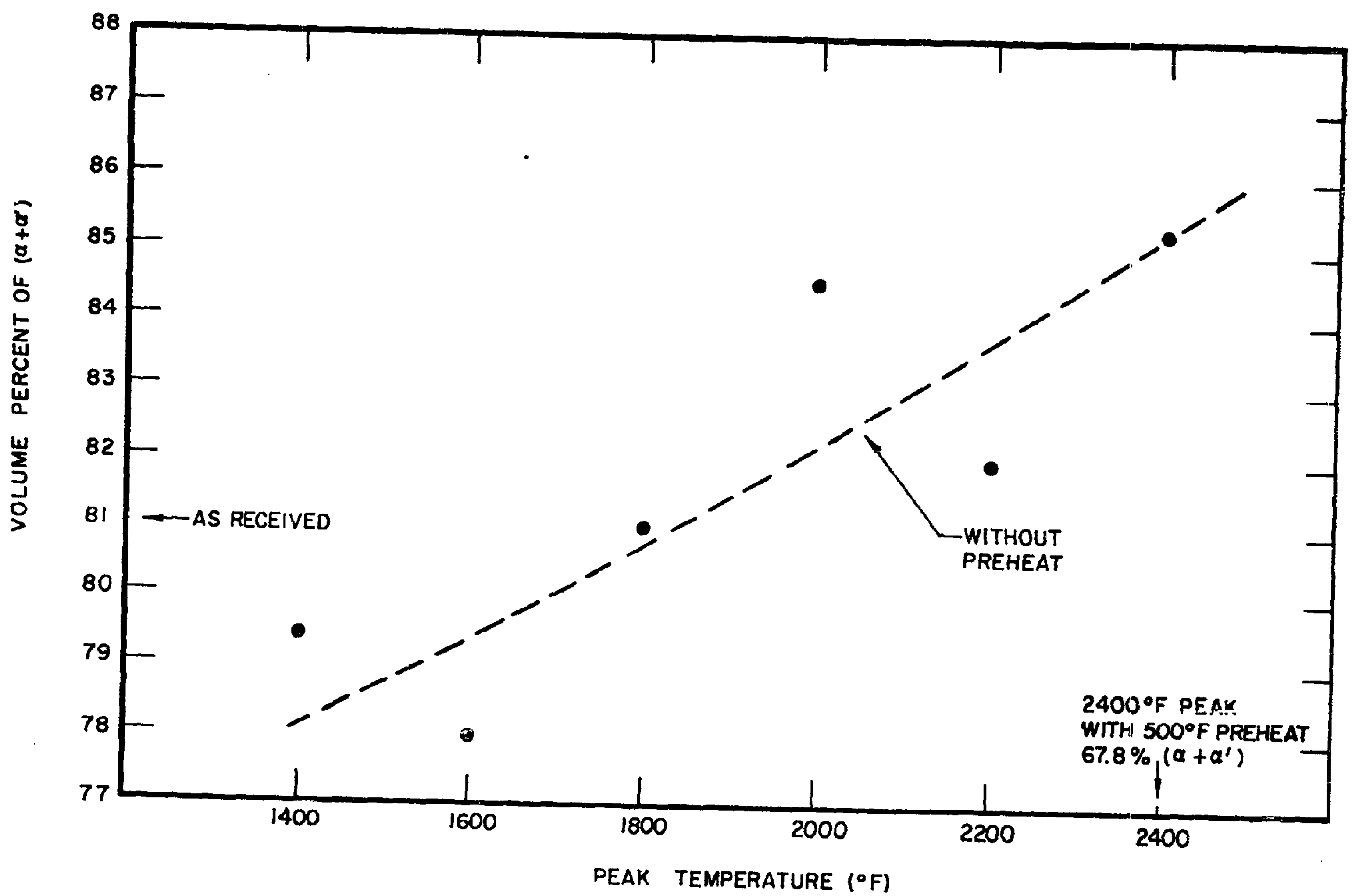


FIGURE 12. VOLUME PERCENT OF $(\alpha + \alpha')$ VS PEAK TEMPERATURE SIMULATED WELD THERMAL CYCLED SPECIMENS

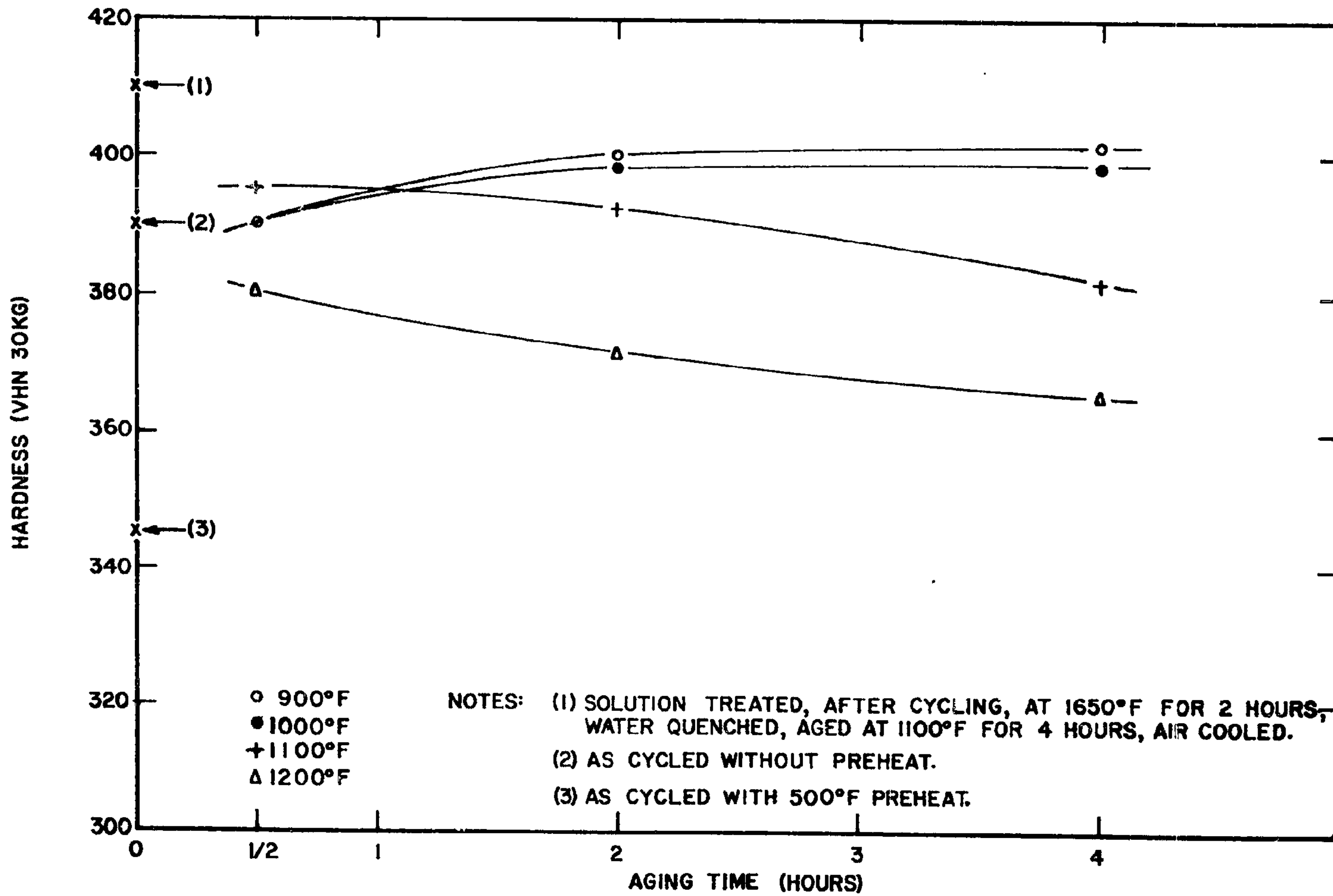


FIGURE 13. EFFECT OF AGING TIME ON HARDNESS FOR VARIOUS AGING TEMPERATURES OF SPECIMENS THERMAL CYCLED to 2400°F PEAK TEMPERATURE WITHOUT PREHEAT

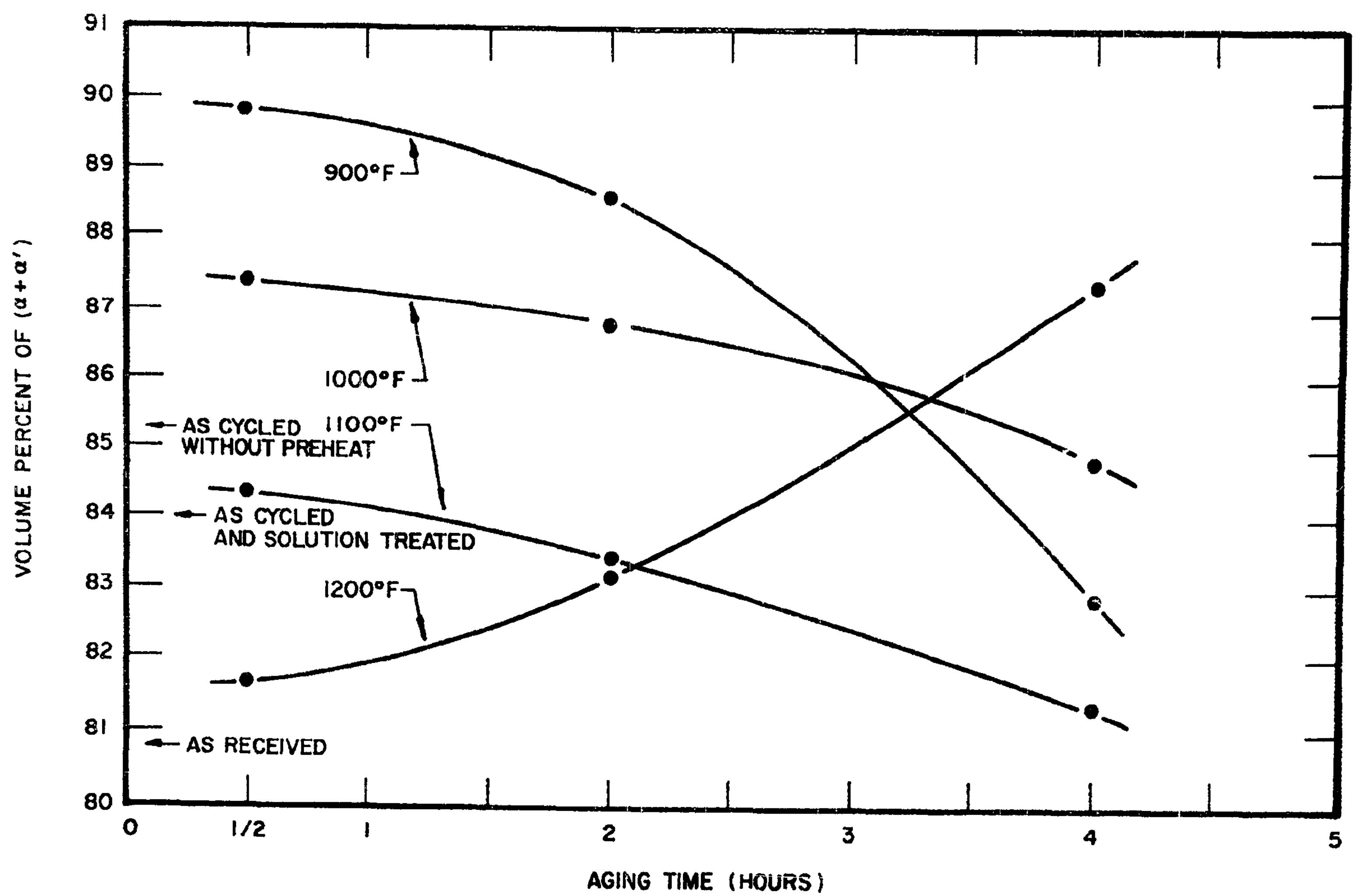


FIGURE 14. VOLUME PERCENT OF $(\alpha + \alpha')$ VS AGING TIME FOR VARIOUS AGING TEMPERATURES OF SPECIMENS THERMAL CYCLED TO 2400°F PEAK TEMPERATURE WITHOUT PREHEAT

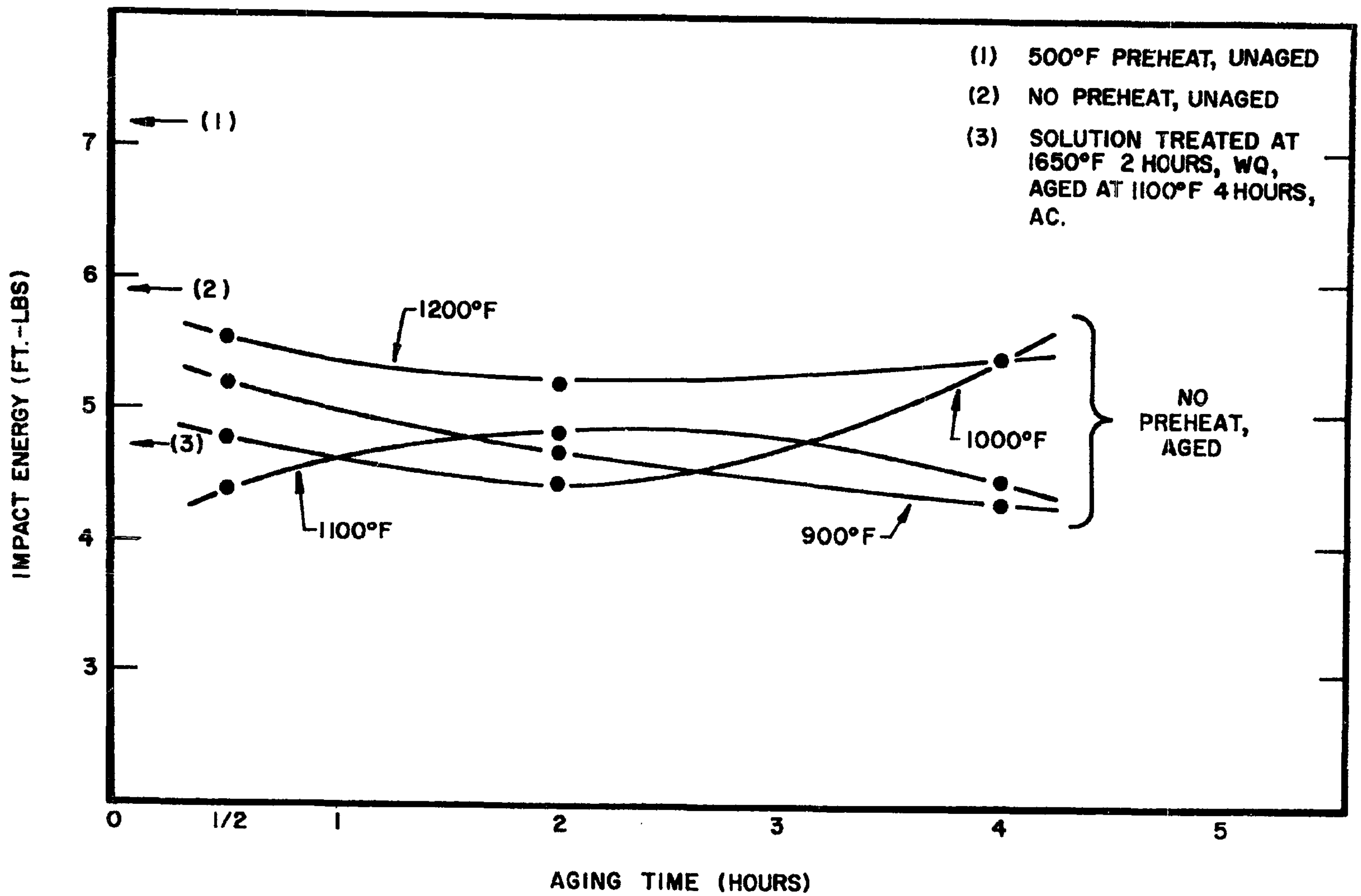
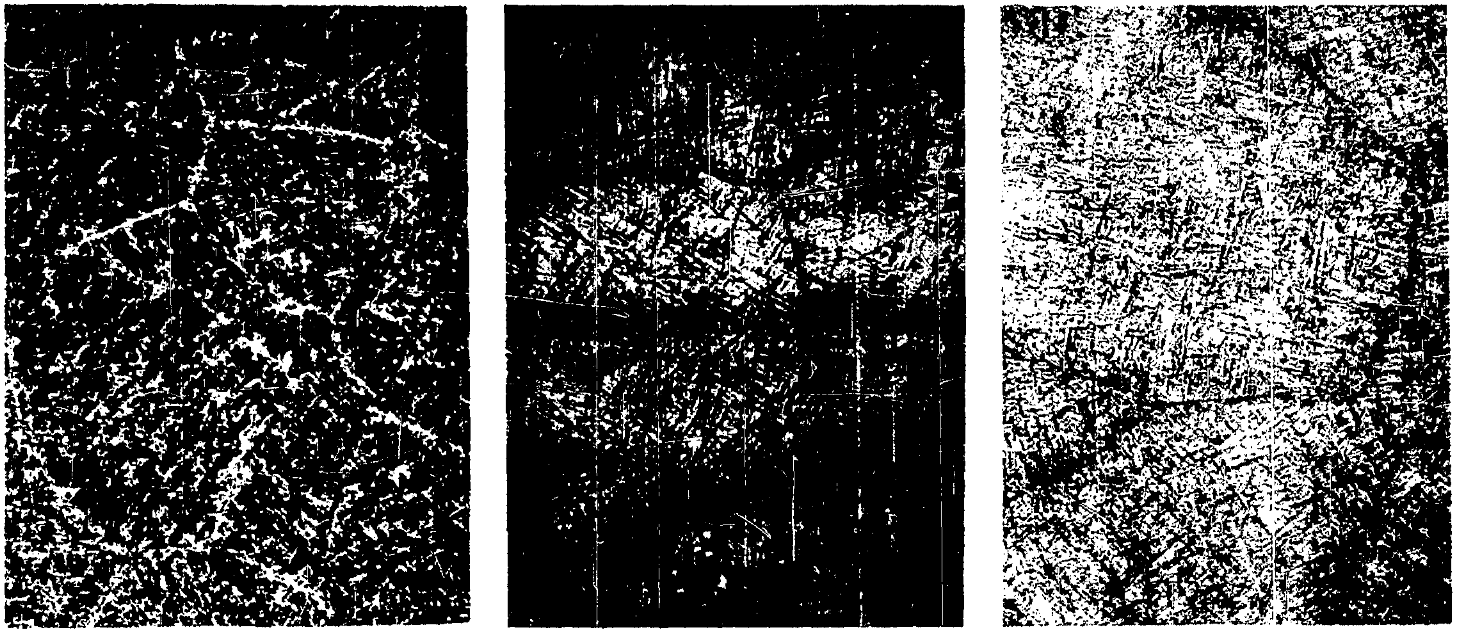
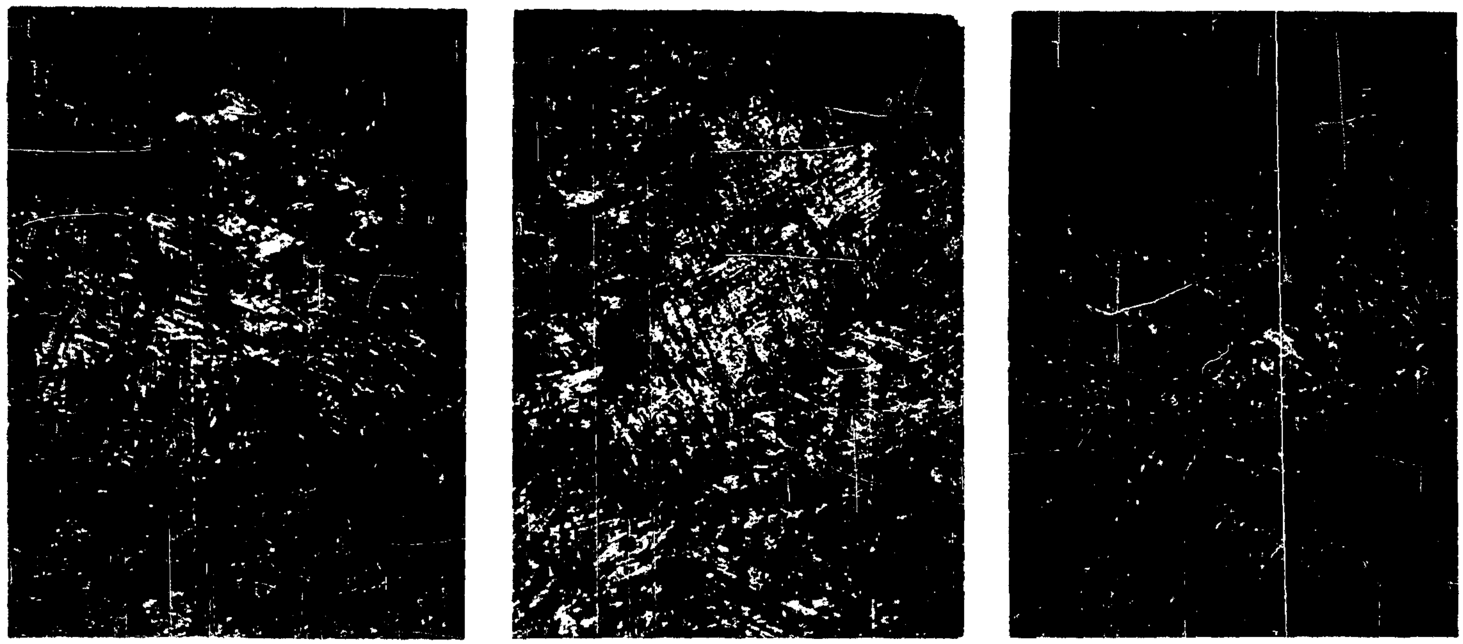


FIGURE 15. EFFECT OF AGING TIME ON CHARPY V-NOTCH IMPACT STRENGTH AT -40°F FOR SPECIMENS THERMAL CYCLED TO 2400°F PEAK TEMPERATURE-AGED AT TEMPERATURES INDICATED



(a) 1/2 Hour　　(b) 2 Hours　　(c) 4 Hours

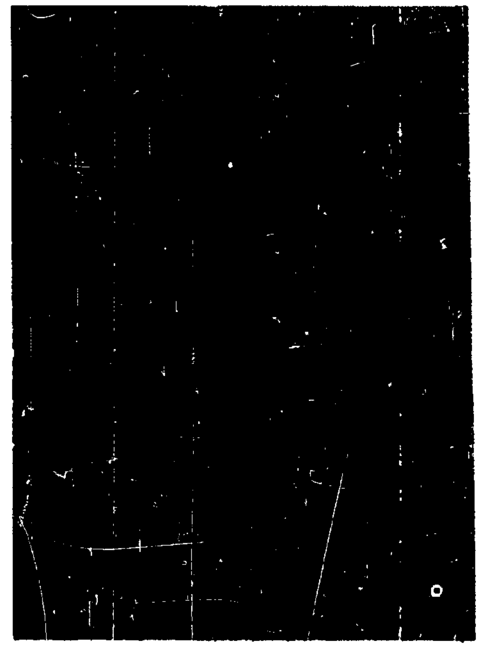
FIGURE 16. THERMAL CYCLED TO 2400°F PEAK TEMPERATURE AGED AT 900°F, AIR COOLED, AGING TIME INDICATED
260X 'R' ETCHANT



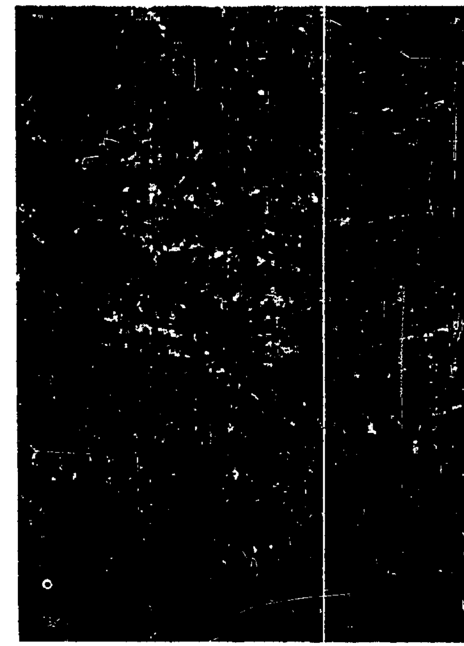
(a) 1/2 Hour (b) 2 Hours (c) 4 Hours

FIGURE 17. THERMAL CYCLED TO 2400°F PEAK TEMPERATURE AGED AT 1000°F, AIR COOLED, AGING TIME INDICATED
260X 'R' ETCHANT

(a) 1/2 Hour

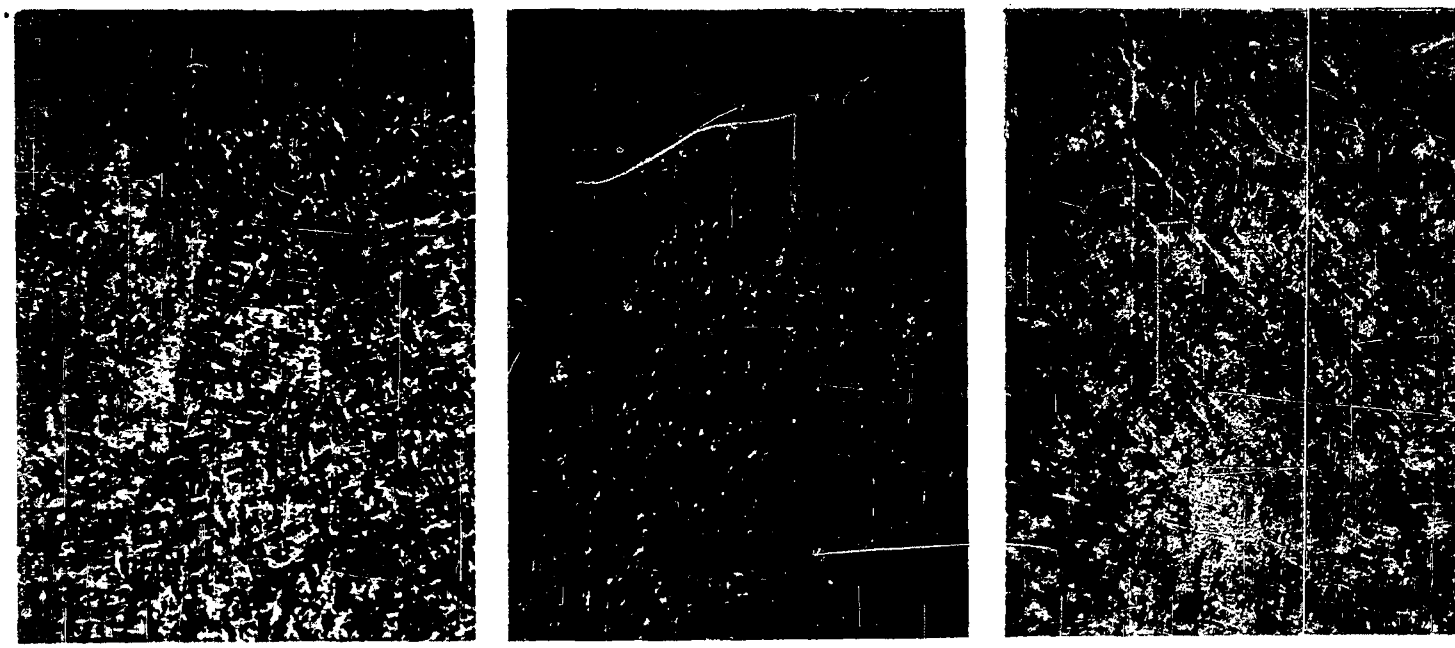


(b) 2 Hours



(c) 4 Hours

FIGURE 18. THERMAL CYCLED TO 2400°F PEAK TEMPERATURE AGED AT 1100°F, AIR COOLED, AGING TIME INDICATED
260X 'R' ETCHANT



(a) 1/2 Hour (b) 2 Hours (c) 4 Hours

FIGURE 19. THERMAL CYCLED TO 2400° PEAK TEMPERATURE AGED AT 1200°F AIR COOLED, AGING TIME INDICATED
260X 'R' ETCHANT

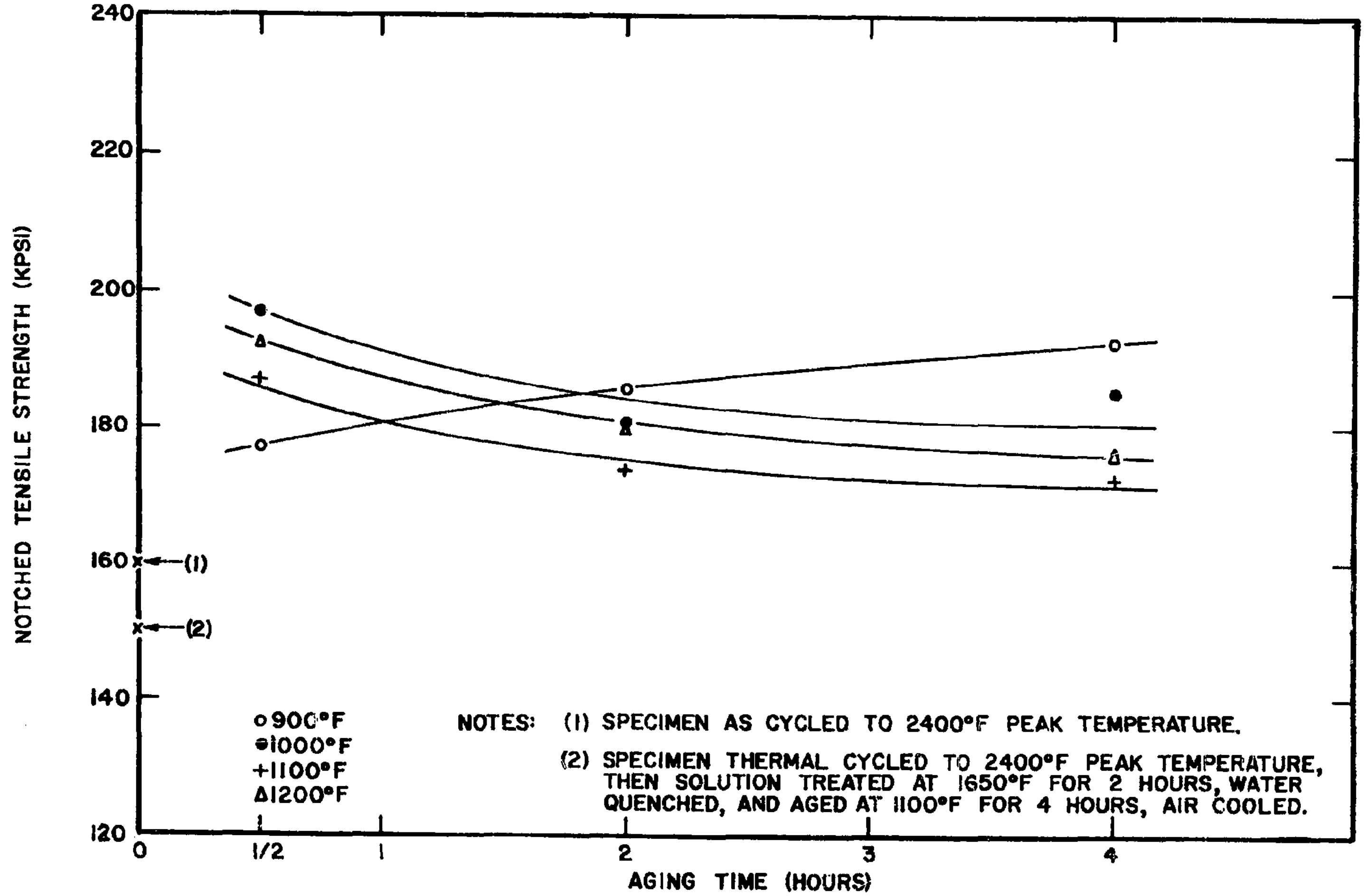
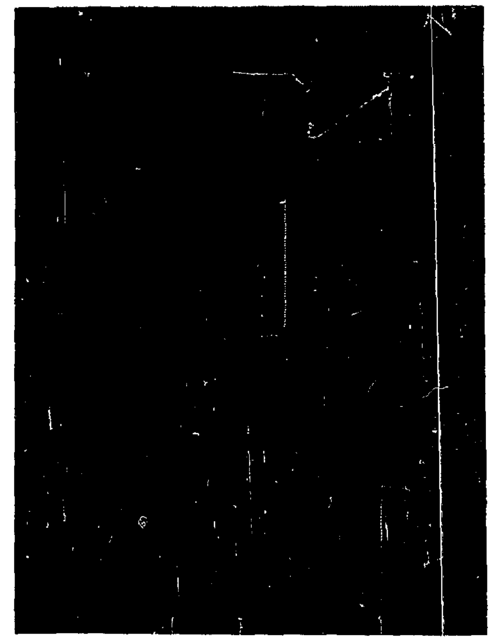


FIGURE 20. EFFECT OF AGING TIME ON NOTCHED TENSILE STRENGTH (K_t = 5.5) FOR VARIOUS AGING TEMPERATURES OF SPECIMENS THERMAL CYCLED TO 2400°F PEAK TEMPERATURE WITHOUT PREHEAT

(a) With 500° Preheat



(b) Without Preheat, Solution Treated at 1650°F 2 Hrs, WQ, Aged at 1100°F 4 Hrs, AC

FIGURE 21. THERMAL CYCLED TO 2400°F PEAK PREHEAT OR POST TREATMENT AS INDICATED
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